Dislocation Mobility in Pure Copper at 4.2 °K[†]

K. M. Jassby* and T. Vreeland, Jr.
W. M. Keck Laboratory of Engineering Materials, California Institute of Technology, Pasadena, California 91109
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Torsional stress pulses of several microseconds duration were applied at 4.2 °K to cylindrical single crystals of copper containing freshly introduced dislocations. Dislocation displacements were measured by means of a double-etch technique, and subsequently the dislocation damping coefficient B was determined to be equal to 0.8×10^{-5} dyn sec/cm^2. While B decreases monotonically with decreasing temperature, the value of B at 4.2 °K is greater than that predicted from theoretical calculations of the interaction between a moving dislocation and the conduction-electron gas in copper.

I. INTRODUCTION

There exists in the literature conflicting experimental evidence in regard to lattice damping of dislocation motion in metals at temperatures below 77 °K. Several investigators[1–4] have reported that lattice damping increases with decreasing temperature in this temperature range and at the lowest temperatures for which measurements were made actually exceeds the magnitude of dislocation damping at room temperature. Other investigators[5,6] have reported only a small residual damping component in the vicinity of 4.2 °K.

Most of the experimental results[1,2,5,6] were derived from conventional internal-friction experiments interpreted in terms of the Granato-Lücke theory.[7] These experiments were carried out at low stress levels with dislocation velocities less than 100 cm/sec. The weakness of this approach is that the distribution of dislocation-loop lengths and total dislocation density, parameters that are important in the Granato-Lücke theory, generally cannot be deduced exactly, and it is difficult to separate the contributions of dislocations of different character. On the other hand, direct stress-pulse experiments[3,4] permitted B to be determined for dislocation velocities greater than one-tenth of the sound velocity in the metal without the qualifying preconditions of any dislocation model.

An extension[8] of the Granato-Lücke theory to explain amplitude-dependent internal friction was employed[9] in one internal-friction technique useful for superconducting metals. In this case the lattice damping of dislocation motion in the normal state of lead was determined in the vicinity of the transition temperature (7.2 °K) without prior estimation of the characteristics of the dislocation network, although the analysis presupposed that phonon damping in the normal state was negligibly small. A damping coefficient was deduced more than one order of magnitude smaller than that obtained from a stress-pulse technique in lead.[3]

In a later work,[10] a second modified-internal-friction technique was reported which employed a low-frequency bias stress and did not require a superconducting transition. This method allowed B to be determined without knowledge of the test crystal's mobile-dislocation density, although assumptions were made concerning the distribution of the dislocation-loop lengths. This technique was applied[10] to an aluminum single crystal and a small temperature-independent damping coefficient was deduced between 4.2 and about 50 °K. This stands in conflict with the large temperature-dependent value of B deduced for fast-moving dislocations from a stress-pulse technique in aluminum.[4]

There exists, at present, no theoretical justification for the large differences in values of B determined in the relatively low-dislocation-velocity internal-friction experiments[9,10] on the one hand, and relatively high-dislocation-velocity (but below the "relativistic" region) stress-pulse experiments on the other. With the aim of providing further evidence concerning the behavior of B in the higher-velocity region, at very low temperatures, the present authors investigated the behavior of freshly introduced dislocations in high-purity cylindrical copper single crystals subjected to torsional-stress impulses of microsecond duration. The experimental technique and results of the investigation are reported below.

II. EXPERIMENTAL TECHNIQUES

A. Test Specimens

A copper single crystal, 3.2 cm in diam with an orientation near a $\langle 111 \rangle$ direction, was grown from 99.999%-pure material by a modified Bridgman technique. This parent crystal was acid cut into sections which were electrolytically machined into cylindrical test crystals 1.3 cm in diam, about 2.5 cm long, and with the cylindrical axis within 0.25° of the [110] direction. Two plane $(\bar{1}1\bar{1})$ surfaces, approximately 2 mm wide, were electrolytically machined on opposite lateral sides of each test crystal, as shown in Fig. 1.

The test crystals were annealed for 2 weeks at 1020 °C in a purified-argon atmosphere. Following this, the crystals were electrolytically polished at − 50 °C to remove traces of evaporation pits. Etching the {111} surfaces of annealed crystals revealed between 10^2 and 10^3 dislocation etch pits per cm^2. The crystals were generally free of substructure.

B. Selectivity Introduced Dislocations

Three sets of {111} slip planes intersect the lateral $(\bar{1}1\bar{1})$ surfaces. Their traces in that surface are given by three ⟨110⟩ Burgers-vector directions, one parallel to the cylindrical axis of the crystal, the remaining two at angles of 60° (Fig. 1). Each $(\bar{1}1\bar{1})$ surface was scratched parallel to one of the latter two directions at intervals of approximately 4 mm along the cylindrical length by means of a thin needle supported in a torsional device. Dislocations were produced on the slip planes whose traces on the surface were parallel to the remaining two Burgers vectors.

C. Application of the Torsional-Stress Pulse

The torsion-testing machine used in previous mobility studies by the present investigators[11, 12] was employed in this experiment. An organic solvent (liquid propane and isopentane in a volumetric ratio of 10 : 1) was used to bond each test crystal

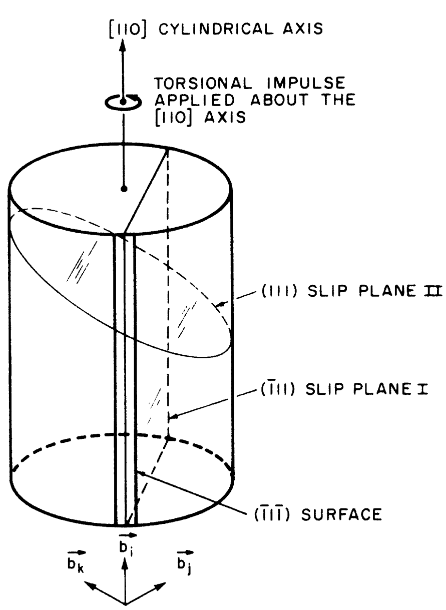

FIG. 1. Geometry of the test specimen employed in the mobility experiment. A $(\bar{1}1\bar{1})$ surface, electropolished on the lateral side of the cylindrical crystal is shown. $\vec{b}_i$, $\vec{b}_j$, and $\vec{b}_k$ are the three Burgers vectors in that surface. The surface is scratched parallel to the $\vec{b}_j$ direction; subsequently it is etched both before and after application of the torsional pulse in order to determine dislocation displacements on the (111) and ($\bar{1}$11) slip planes.

to the steel torsion rod. The solvent hardened at about 44 °K but differential thermal contraction on cooling below this temperature did not cause significant plastic deformation in the test crystal at the crystal: steel-rod interface. The torsion pulse was applied as in previous investigations. At any horizontal cross section of the test crystal, the time between arrival of the torsional loading wave and that of the reflected unloading wave was equal to twice the distance from that cross section to the free end of the crystal divided by the torsional-wave velocity. The state of torsional strain at the interface between the steel torsion rod and test crystal was monitored by a set of strain gages attached to the cylindrical surface of the torsion rod several cm above the test crystal. Subsequently, the time history of the state of torsional stress at any horizontal cross section could be calculated by means of simple numerical techniques.

D. Resolved Shear Stress on Mobile Dislocations

With the crystal geometry prescribed here, the mobile dislocations introduced by means of the needle scratches could contain any one of five possible Burgers vectors. We assume that dislocations meet the slip-plane: free-surface interaction at 90°, and then it can be seen from purely geometrical considerations that near the lateral free surface the dislocations on any particular set of slip planes had two possible orientations, either edge (one case) or 30° mixed (two cases). With the condition of a pure torsional-stress state, the stress resolving factors for the various dislocations are tabulated in Table I.

III. EXPERIMENTAL RESULTS

A. Dislocation Displacement Measured as a Function of Torsional Impulse

Four crystals were tested at three different values of applied torque (equal to 1.2×10^6 dyn cm for two tests, 2.4×10^6 dyn cm for one test, and 3.1×10^6 dyn cm for one test). A standard double-etch technique was used to reveal dislocation positions before and after application of the torsional-stress pulse (Fig. 2). For the two tests at the higher levels of applied torque, dislocation motion away from a scratch was generally uniform along the scratch. In the case where dislocation displacements were relatively large (≥ 0.02 cm, say), it was difficult to determine along which of the two possible slip directions (Fig. 2) the displaced dislocations had moved. However, as the two slip directions formed an equilateral triangle with the scratch direction, the displacement of the furthest moving dislocation at any point along the scratch was simply given by $(2/\sqrt{3})S_\perp$, $S_\perp$ being the perpendicular distance from the scratch to the dislo-

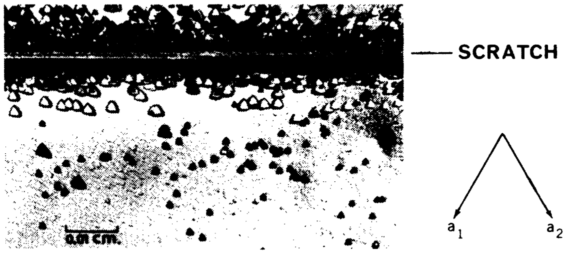

FIG. 2. Area on a latteral $(\bar{1}1\bar{1})$ surface of a copper torsional specimen. The surface has been scratched by means of a thin needle, etched, tested in the torsion machine at 4.2 °K, then finally re-etched. The small triangular pits indicate the positions of dislocations displaced from the scratched region by the applied-stress pulse. The directions a_1 and a_2 indicate the two possible slip directions for dislocations produced by the needle scratch.

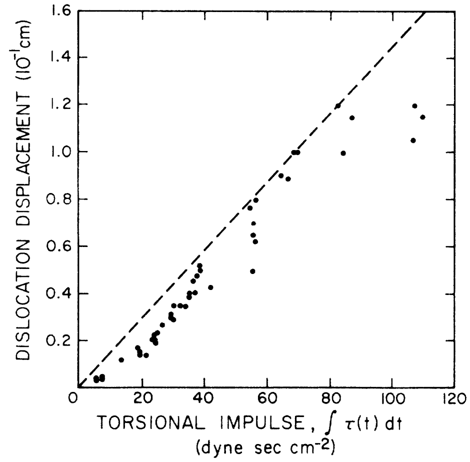

FIG. 3. Dislocation displacement as a function of total torsional impulse, for 30° mixed dislocations in copper at 4.2 °K.

cation. The density of grown-in dislocations was considerably less than that of the fresh ones displaced from a scratch. Consequently, because of the high-dislocation-density change across the displaced dislocation front, it was possible to make a lower-bound estimate of the displacements of the furthest moving dislocations.

In a few isolated instances, dislocations were displaced in slip bands from the needle scratches, thus indicating the presence of operative Frank-Read sources. Here the slip direction was self-evident, and employing standard Berg-Barrett x-ray diffraction techniques, the operative Burgers vector was determined. In the slip bands, the dislocations were always of mixed character, with the Burgers vector at 30° to the dislocation line ($\vec{b}_2$, $\vec{b}_3$, and $\vec{b}_5$ of Table I).

For the two crystals tested at the lowest level of applied torque, dislocations introduced by the needle scratches near the free end of the crystals (subject to the smallest torsional impulse) were considerably less mobile. For many of these dislocations there was no measureable displacement upon application of the torsional pulse.

In each of the four tests, the average distances from the free end of the crystal were measured for the dislocations displaced farthest from each scratch. The torsional impulse for each of these dislocations $\int \tau(t)\,dt$, was evaluated, where the torsional stress $\tau(t)$ was taken as that at the horizontal cross section (see Sec. II C) corresponding to the average distance of the dislocation from the free end of the crystal. The maximum measured dislocation displacement is plotted as a function of torsional impulse in Fig. 3. Slip-band measurements were in accord with those of individual dislocations. The maximum stress-pulse time (far from the free end of a crystal) was of the order of 10 μsec. Hence, maximum dislocation velocities (at the highest value of applied torque) exceeded 10^4 cm/sec.

B. Character of the Displaced Dislocations

In this experiment the individual Burgers vectors of the displaced dislocations were not determined except, as explained in Sec. III A, in the few instances in which dislocations moved in slip bands. These latter dislocations were 30° mixed in character. Referring again to Table I, we note that for a given scratch, dislocations of this type were the only dislocations with the same stress resolving factor on both operative slip planes I and II. As the experimental results displayed in Fig. 3 demonstrate a reasonable correlation between displacement and torsional impulse over a wide range of values of both parameters, and as the isolated slip-band measurements agreed well with the data from individually displaced dislocations, it was

TABLE I. Torsional-stress resolving factors for mobile dislocations produced by needle scratches. Slip plane I refers to the slip plane parallel to the cylindrical axis of the crystal while slip plane II refers to the slip plane whose normal vector is 35° 16′ from the axis.

	$\vec{b}_1$ (edge)	$\vec{b}_2$ (30° mixed)	$\vec{b}_3$ (30° mixed)	$\vec{b}_4$ (edge)	$\vec{b}_5$ (30° mixed)
Slip plane I	0.94	0.47	0.47		
Slip plane II			0	0.47	0.47

concluded that the accumulated data of Fig. 3 referred to the mobility of 30° mixed dislocations.

C. Calculation of the Dislocation Damping Coefficient B

The relation between dislocation displacement d and dislocation velocity v is

$$d = \int v(t)\, dt. \tag{1}$$

The operative resolved shear stress on the mobile dislocations τ_R is equal to 0.47τ. Then from Fig. 3,

$$d = K \int \tau_R(t)\, dt, \tag{2}$$

where $K = 3.1 \times 10^{-3}$ cm^3/dyn sec is the asymptotic slope through the experimental data. Comparing Eq. (1) with Eq. (2), and exploiting the expression for the damping coefficient ($Bv = \tau_R b$), we find

$$B = b/K \approx 0.8 \times 10^{-5} \text{ dyn sec/cm}^2, \tag{3}$$

where $b = 2.55 \times 10^{-8}$ cm for copper. In Fig. 4, we have plotted B as a function of temperature as determined in this experiment and in previous direct experiments[11,12] on copper.

IV. DISCUSSION OF RESULTS

A. Extrinsic Drag Forces

The dislocations introduced into the test crystals by the needle scratches and subjected to relatively small torsional impulses displayed a reduced mobility (see Sec. III A and Fig. 3). We attribute this effect to the influence of extrinsic forces characterized by dislocation-pinning points such as nonglissile jogs and nonmobile dislocation junctions introduced by the needle scratches. The initial lengths and shapes of the fresh dislocations were not identified (with etch-pit techniques, only the surface characteristics may be studied), but we can assume that the resolved shear stress must be of sufficient strength and applied for a period long enough to enable the dislocations to free themselves from their pinning points in the usual way. It has been predicted[13] that at stress levels near the minimum stress required for dislocation breakaway and for periods corresponding to the minimum time for breakaway, the relation between dislocation velocity and resolved shear stress is markedly nonlinear even when an intrinsic viscous damping mechanism is operative. This effect has been confirmed in recent mobility experiments in zinc[14] and the present experimental results also indicate the influence of extrinsic forces at the lower levels of torsional-stress pulse.

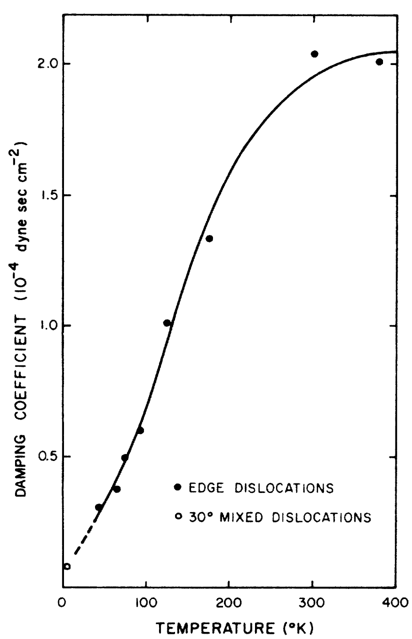

FIG. 4. Damping coefficient as a function of temperature for dislocations in copper.

B. Comparison with Previous Experimental Work

The dislocation-damping coefficient in copper at temperatures below 77 °K has been estimated in two internal-friction experiments. In the former work,[5] a decreasing value for B was found with decreasing temperature down to 20 °K (the lowest temperature for which data was taken), and B was estimated to be about 2×10^{-5} dyn sec/cm^2 at this temperature. One would expect, from this data, a value of B of the order of 10^{-5} dyn sec/cm^2 at 4.2 °K. In the latter work,[6,15] while qualitatively a similar temperature dependence was observed, B was estimated to be of the order of 5×10^{-7} dyn sec/cm^2 at 4.2 °K, or less than one-tenth of the present results.

Additional experimental evidence pertaining to the magnitude of the intrinsic lattice damping of dislocation motion at 4.2 °K in metals with a crystal structure similar to that of copper has been obtained for aluminum[10] and lead in the normal state[9] in two modified-internal-friction techniques. In both cases, B was deduced to be of the order of 10^{-5} dyn sec/cm^2 in magnitude at 4.2 °K. In both these materials, B displayed a temperature dependence qualitatively similar to that found in the internal-friction experiments in copper.

The present experimental results in copper, obtained from direct mobility measurements, confirm the accumulated experimental evidence indicating that B decreases as a function of decreasing temperature. In addition, they support a value for B of the order of 10^{-5} dyn sec/cm^2 at 4.2 °K. These results stand in contrast to the only other direct measurements[3,4] of dislocation mobility near 4.2 °K in metals with close-packed structures, these latter exhibiting the opposite temperature dependence of B and a value for B at 4.2 °K at least one order of magnitude greater than the present results.

C. Comparison with Theoretical Predictions

It is generally believed that in the liquid-helium temperature range conduction electrons provide the dominant contribution to energy absorption from a uniformly moving dislocation. The interaction between electrons and moving dislocations has been treated independently by several investigators.[16–20]

The early treatment of Mason[16] employed the properties of electronic viscosity, but Kravchenko[17] and others demonstrated that electronic viscosity was inapplicable to moving dislocations at low temperatures. Kravchenko[17] and Holstein[18] and later Huffman and Louat[19] considered separately the interaction between the conduction-electron gas and individual Fourier components of a moving dislocation's velocity field. The total rate of energy absorption by the electron gas was found by integrating over all Fourier components. Brailsford[20] pointed out an error in the treatment of Huffman and Louat[19] and verified the essential correctness of the form of Holstein's calculation.

Holstein considered only the interaction between the electron gas and the dilatational field of a moving dislocation. Brailsford indicated that this was sufficient, as the Fourier components representing the shearing field of a moving dislocation interact weakly with the electron gas. (This may readily be verified by a treatment similar to that of Ref. 20.) It appears, then, that only the edge component of a moving dislocation in copper is influenced strongly by the electron gas.

Brailsford's calculation for the electronic-damping coefficient is[20]

$$B=\left(\frac{1-2\nu}{1-\nu}\right)^2 \frac{n_0 m v_F b_e^2}{96}\; q_D\varphi\left(x=\frac{q_D}{q_{\mathrm{TF}}}\right) , \qquad (4)$$

where $\varphi(x)$ is defined by

$$\varphi(x)=\tfrac{1}{2}[(1+x^2)^{-1}+x^{-1}\tan^{-1}x] . \qquad (5)$$

The various parameters of Eq. (4) are listed for copper: ν, Poisson's ratio $=0.32$; n_0, the number of conduction electrons per unit volume $=8.5\times10^{22}$ cm^{-3}; m, the electronic mass $=9.1\times10^{-28}$ g; v_F, the Fermi velocity $=1.6\times10^8$ cm/sec; b_e, the magnitude of the edge component of the dislocation Burgers vector $=\frac{1}{2}b$; q_D, the Debye wave vector $=1.7\times10^8$ cm^{-1}; q_{TF}, the reciprocal of the Thomas-Fermi screening length $=1.8\times10^8$ cm^{-1}.

The various parameters, when substituted into Eq. (4), give $B\approx0.07\times10^{-5}$ dyn sec/cm^2. This value is less than one-tenth of the magnitude for B determined in the present experiment. It is suggested that the analytical considerations be further tested experimentally by a study of the mobility of dislocations of orientations other than the 30° mixed considered here.

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*Present address: School of Engineering, Tel Aviv University, Ramat Aviv, Tel Aviv, Israel.

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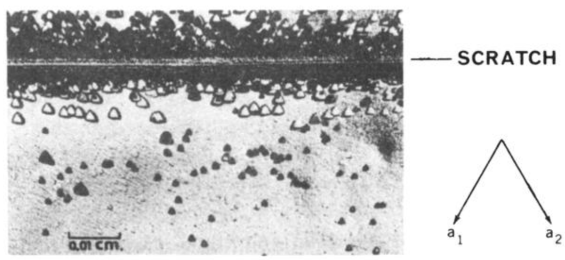

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